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(54) PRODUCTION METHODS FOR STEEL SHEET

(71) We, NIPPON STEEL CORPORATION, a Japanese Company of No. 6-3, 2-chome, Ote-machi, Chiyoda-ku, Tokyo, Japan, do hereby declare the invention, for which we pray that a patent may be granted to us, and the method by which it is to be performed, to be particularly described in and by the following statement:-

This invention relates to methods of producing a steel sheet, and in particular to methods of producing steel sheet containing molybdenum and suitable for the manufacture of pipe lines and fittings therefor. The invention further relates to such sheet steel itself. Steel sheet produced in accordance with this invention may display excellent DWTT (drop weight tear test) characteristics at low temperatures, as specified by API SR6, such as at - 30°C or lower. In this Specification, the term 'sheet' is used to refer to products such as strip, plate and the like, as well as strict sheet products.

In recent years, greater importance has been placed on natural gas as a new energy source and prospecting and exploration is being undertaken in Arctic regions to seek new gas fields. For the prospecting and exploration, demands are made for high-tensile, high toughness, large-diameter gas line pipes and fittings which can realize efficient and economical transportation of gas to consumer cities.

Steel pipes for use in gas pipe lines are required to have excellent ductility, as determined by DWTT which represents the property of preventing brittle fracture, as well as an excellent Charpy impact value in order to prevent a large scale ductile fracture of the pipe lines and the fittings. Steel sheets for making gas line pipes which satisfy the above severe material properties have been produced by the so-called "controlled rolling method" (hereinafter called "CR"), and Nb-containing steels (hereinafter called "Nb-steels") have mainly been used for this purpose.

Nb-steel is one of the most commonly used steel grades and has excellent properties, but on the other hand this steel lacks the following characteristics:

(1) In order to utilize Nb effectively for precipitation hardening and refinement of grains, it is necessary fully to dissolve into solid solution the coarse precipitation of Nb(CN) contained in the steel slab when heating the slab prior to hot rolling. However, precipitated Nb(CN) is stable at high temperatures so that it is not fully dissolved into solid solution at less than 1150°C, and it is necessary to maintain a considerably long holding time for the heating,

thus lowering the productivity of the heating furnace.

(2) When the steel slab is heated to a temperature up to 1150°C, where Nb(CN) begins to dissolve into solid solution, the amount of Nb in solid solution thus obtained varies considerably due to fluctuations in the heating temperature. When the amount of Nb in solid solution increases excessively the austenite grains (heated γ grains) formed during the heating take the form of considerably mixed grains so that the toughness deteriorates. Even if the rolling is performed under the same conditions as in the case of a steel slab in which the austenite grains are not mixed, the strength increases excessively so that the material quality lacks stability.

(3) Nb is an element which strongly inhibits recrystallization of the rolled austenite grains (rolled γ grains) during the rolling, so that below about 1050°C no satisfactory recrystallization occurs. Therefore, recrystallization of the elongated austenite grains does not take place before the grains are converted into fine recrystallized rolled austenite grains during the rolling, leading to difficulties such as the reduction amount is not enough in the non-recrystallization temperature zone, and when the rolling is finished at a high temperature range in the non-recrystallization zone, the rolled structure thus obtained has a coarse mixed grain structure and is susceptible to the occurrence of Widmanstatten structure, particularly in the case where the final plate is relatively thick.

(4) When the degree of warm rolling is increased, the yield ratio

$$YR\ (YR = \frac{\text{yield strength (YS)}}{\text{tensile strength (TS)}})$$

can become as high as 95%, so that the production of steel pipes such as by the UO process becomes difficult and a deterioration in the yield strength due to the Bauschinger effect is considerable: an excessive yield strength is thus required for the steel sheet.

(5) During welding of the steel sheet, precipitated Nb(CN) is apt to resolidify and thus the hardness of the sheet increases greatly, and the toughness of the weld metal and the welding heat-affected zone (HAZ) deteriorates considerably. Also when the stress-relieve annealing (SR) is performed, the Nb which resolidifies during the welding precipitates to lower the toughness remarkably.

(6) When a continuous casting process (CC) is used for the production of steel slabs, Nb(CN) precipitates at the grain boundaries of the austenite grains to cause intergranular embrittlement, which leads to surface cracking of the steel slab.

Extensive studies over many years have been made for the development of a steel composition which overcomes the above defects of conventional Nb-steels and which still has the advantageous precipitation hardening property and the refinement of grains achieved by conventional Nb-steels. It has been found that the addition of a very small amount of Mo is most effective for this purpose. However, it has also been found that some molybdenum-containing steel compositions show severe embrittlement when subjected to a warm rolling under certain rolling conditions. This invention stems from research on the rolling conditions which cause the above embrittlement.

According to this invention, in a method of producing a steel sheet (as defined herein), there is provided the steps of heating a steel slab containing (by weight, as are all percentage-contents mentioned herein unless otherwise specified) 0.01 to 0.13% C, 0.05 to 0.8% Si, 0.8 to 1.8% Mn, 0.01 to 0.08% total Al, 0.08 to 0.40% Mo, and not more than 0.015% S with the balance being iron and unavoidable impurities to a temperature not higher than 1150°C, and hot rolling the heated steel slab, the hot rolling including at least three passes with a minimum reduction percentage of not less than 2% in each rolling pass within the temperature range of 900 to 1050°C, and the total hot rolling reduction percentage at 900°C or lower being not less than 50%, and the finishing temperature of the hot rolling being not higher than 820°C.

The steel slab composition may be modified so as further to contain at least one of 0.02 to 0.20% V, 0.001 to 0.03% REM (rare earth metals), 0.0005 to 0.03% Ca, 0.004 to 0.03% Ti, not more than 0.6% Cr, not more than 0.6% Cu and not more than 2.5% Ni, and also may be modified so that the nitrogen content is limited to the range of 0.001 to 0.009% when Ti is added, and satisfying the REM/S ratio of 1.0 to 6.0 when REM is contained.

The invention extends to steel sheet whenever produced in accordance with a method of this invention.

In order that the invention may better be understood, it will now be described in greater detail, reference being made to the accompanying drawings, in which:

Figure 1 is a graph showing the effect of the molybdenum content on recrystallized rolled austenite grains and vTrs values;

Figure 2 is a graph showing the relation between the heating temperature and the heated austenite grain size when steel B (Table 1) is heated to various temperatures and held thereat for 60 minutes;

Figure 3 is a graph showing the relation between the rolling temperature and the rolled

austenite grain size under certain rolling conditions;

Figure 4 is a graph showing the relation between the number of rolling passes in the temperature range of 950 to 980°C and the rolled austenite grain size;

Figure 5 is a graph showing the relation between the reduction amount at a temperature not higher than 900°C and the yield point and DWTT 85%SATT value in steel C (Table 1);

Figure 6 is a graph showing the relation between the finishing temperature and both the yield point and the DWTT 85% SATT values in steel C (Table 1);

Figure 7 shows shape and size of a test piece for DWTT (the drop weight tear test according to API); and

Figure 8 illustrates how the fracture of the test piece is observed.

It has been found that an addition of Mo in very small amounts to a steel increases the tensile strength (TS) and the yield strength (YS) due to its hardening-improvement effect, lowers the yield ratio (YR), and under certain conditions in the high-temperature zone during the rolling it is most effective at refining the recrystallized rolled austenite grains, while in the temperature range below 900°C it is effective to elongate the rolled austenite grains and to refine the rolled structure in a similar manner to Nb and V. It should be noted particularly in this connection that the recrystallization-preventing characteristic of Mo is less strong that that of Nb but is stronger than that of V, depending on the amount of Mo addition and the heating and rolling conditions.

It follows that from the above excellent properties of Mo, the re-crystallized rolled austenite grains can be refined more easily in a Mo-containing steel than in a Nb-steel, and the re-crystallized rolled austenite grains can be elongated by rolling at 900°C or lower in a Mo-containing steel so that a very fine rolled structure with considerably less mixed grains can be achieved. Also, a Mo-containing steel has the advantage over V-steels in that Mo, unlike V, is most effective for refining the rolled austenite grains in the high-temperature zone, and that the rolled austenite grains can be elongated. Hence a fine rolled structure can be achieved even if the rolling is not performed at so low a temperature, because Mo is better than V at preventing recrystallization.

Further advantages which may arise from this invention can include the following:

(1) The heating problem inherent with Nb-steels does not occur because no Nb is contained, and a very stable balance can be obtained between strength and toughness;

(2) The steel composition is suitable for continuous casting and if this process is used the problem of surface cracking does not occur; and

(3) The yield ratio (YR) is 2 - 10% lower than that of the Nb-steels, depending on the Mo content (although influenced by contents of C and Mn) so that pipe manufacturing such as by the UO forming process can easily be performed, especially since a deterioration in the yield strength (YS) due to the Bauschinger effect is less or the yield strength can increase with some steel compositions.

In order to make full use of the possible merits of Mo, as mentioned above, and in order to achieve properties suitable for the pipe line steel sheets - i.e. the strength, toughness and weldability of the steel sheet as a base material, and the toughness and resistance to hydrogen cracking of the welded portion - it is essential that the Mo content satisfies the specified range.

Figure 1 shows the relation between the Mo content in a steel and the grain size of rolled austenite grains, the steel containing from 0.04 to 0.11%C and from 1.08 to 1.52%Mn. It is clear from the graph that with a Mo content of less than 0.08%, there is no practical effect on the refinement of the rolled austenite grains and thus a Mo content of at least 0.08% is necessary for this purpose. On the other hand, with a Mo content exceeding 0.40% a large amount of bainite or island martensite structure is produced in the rolled structure, although the rolled austenite grains are considerably refined, so that a deterioration in the toughness occurs and the resistance to hydrogen cracking deteriorates, in spite of the increase in the tensile strength. Thus, the upper limite of the Mo content is set at 0.40%.

Regarding the recrystallization preventing effect of Mo, it has been revealed by the studies and experiments that the recrystallization temperature increases as the Mo content increases, but with 0.08% Mo content the rolled austenite grains are elongated by rolling at or below 900°C and thus this level of Mo content is effective to refine the rolled structure. Therefore, the range of 0.08 to 0.40% of Mo is desirable.

As described above, in order to make full use of the advantages of Mo-steels, this invention defines the heating and rolling conditions for the production of steel sheet.

It has been found through the extensive studies that the austenite grains become coarse once the rolling is done with a light reduction of less than 2% in the temperature range of from 1050 to 900°C, so that the total effect of the subsequent high-reduction rolling passes are reduced by almost half and the refinement of the austenite grains hardly occurs, thus failing to obtain a high-toughness final product. It has further been found that if three or more rolling passes, each with a reduction exceeding 5%, are given in the temperature range of from 1050 to 900°C, the recrystallized grains are refined still further so that the rolled austenite grains

are refined still further so that the rolled austenite grains are refined with improvements in the DWTT property.

The reduction amount, used in this Specification, is defined as follows. If the thickness of the steel before reduction is H and the thickness after reduction is h, the reduction amount (%) is:

$$\text{Reduction amount (\%)} = \frac{H - h}{H} \times 100$$

In Mo-steels, the embrittlement phenomena cannot be eliminated and a satisfactory low-temperature toughness cannot be assumed unless the rolled austenite grains are refined and elongated giving a refined rolled structure. For this reason, it is necessary to reduce the size of the heated austenite grains to be as small as possible.

Figure 2 is a graph showing the relation between the heating temperature and the grain size of the heated austenite grains, and it is clear from this graph that the heating should be done at a temperature not higher than 1150°C and preferably in the range from 1050 to 1150°C. In view of the possible coarsening of the heated austenite grains due to a lengthening of the holding time during the heating, it is desirable that the holding time is 2 hours or less.

It is also necessary to refine further the heated austenite grains refined by the rolling in the recrystallization zone, so the grains become finer rolled austenite grains (typically not less than ASTM No. 6).

Figure 3 is a graph showing the relation between the rolling temperature - using the same rolling conditions - and the grain size of the rolled austenite grains. It can be clearly understood from the graph that when the rolling is done in the temperature range of from 1050 to 900°C, the size of the obtained rolled austenite grains is equal to or finer than ASTM No. 6. Therefore, the rolling temperature in the recrystallization zone is preferably from 1050 to 900°C. It is quite acceptable for the rolling to be done first at a temperature above 1050°C and then in a temperature range of from 1050 to 900°C.

Figure 4 is a graph showing the relation between the number of rolling passes - under the same rolling condition - and the grain size of the rolled austenite grains so obtained.

It is clear from the graph that no satisfactory refinement of the recrystallized rolled austenite grains can be achieved, unless at least three rolling passes are given. Also, regarding the reduction percentage per rolling pass in the temperature range of from 1050 to 900°C, it has been revealed that the effect of the reduction percentage on the grain size of the recrystallized rolled austenite grains is generally small with Mo-steels, but when the minimum reduction in the above temperature range is less than 2%, the hot deformation of the austenite grains is not enough and the grains which have coarsened after the reduction cannot subsequently be refined no matter how large the subsequent reduction.

From the above, it follows that in the method of this invention, at least three reductions are necessary, each with a reduction percentage exceeding 2% in the temperature range of from 1050 to 900°C.

It is also necessary to refine the elongated rolled structure by rolling the fine recrystallized rolled austenite grains in the non-recrystallization temperature zone of not higher than 900°C. For this purpose, the total hot rolling reduction percentage must not be less than 50%. When the total reduction percentage at 900°C or lower (ie. the non-recrystallization zone) is 50% or more, the yield point and toughness are considerably improved as will be appreciated from Figure 5. If the total hot rolling reduction percentage is less than 50%, it is not possible to maintain the transition temperature of 85% brittle fracture characteristic in the drop weight tear test (DWTT 85% SATT) at -30°C, which characteristic is important for steels to be used in pipe lines. Even when the total reduction percentage at 900°C or lower is not less than 50%, only a poor DWTT property is obtained and sufficient strength is not achieved if the finishing temperature is 820°C or higher, as shown by Figure 6.

On the basis of the above, the hot rolling conditions in the non-recrystallization zone are defined in the present invention as that a total reduction of not less than 50% is given at a temperature of not higher than 900°C. and the finishing temperature of the hot rolling is not higher than 820°C.

Regarding the rolling temperature immediately before, or several passes before, the finishing rolling pass, it has been confirmed through experiments that a good low-temperature toughness can be achieved when the temperature of the last few passes is below the Ar_3 transformation point, if the steel composition and the rolling conditions are within the scope of the present invention. Therefore, rolling partially in the dual-phase (γ - α) zone is within the scope of the present invention.

It should be also understood that the steel sheet, after rolling, may be heated to a temperature not higher than the AC_1 point and cooled for the purpose of dehydrogenation, and so on. In this case, any island martensite, and so on, is decomposed to cementite and the

yield point increases, while the tensile strength is lowered to improve the toughness, and also the resistance to hydrogen cracking is improved. Therefore, especially for thick plates, such heating can be most advantageous.

The reasons for the limitations placed on the constituents of the steel composition used in the method of this will now be discussed.

The basic steel composition used in this invention consists of:-

C: 0.01 to 0.13%,
Si: 0.05 to 0.8%,
Mn: 0.8 to 1.80%,
Total Al: 0.01 to 0.08%,
S: not more than 0.015%,
Mo: 0.08 to 0.40%, and
Fe and unavoidable impurities: balance.

The lower limit of the carbon content is defined by the minimum amount required for refinement of the base steel structure and for assuring sufficient strength in a welded portion, as well as for assuring that carbide-forming elements, such as V, can fully exert their effects. On the other hand, when the carbon content is excessively large, a large amount of bainite and island martensite is formed, even with Mo contents within the range from 0.08 to 0.40%, to have an adverse effect on the toughness and to lower the weldability. Thus the upper limit of the carbon content is set at 0.13%. In order to eliminate the adverse effects on the toughness of the segregation zone, not more than 0.1% of carbon should be contained.

Silicon inevitably is contained as a deoxidizing agent in the steel and with a silicon content of less than 0.05%, the toughness of the base steel deteriorates: and this amount is thus set as the lower limit for the silicon content. On the other hand, an excessive silicon content has an adverse effect on the cleanliness of the steel and therefore the upper limit of the silicon content is set at 0.8%.

Manganese is an important element for maintaining the required strength and toughness of a low-carbon steel, such as is used in this invention. With a manganese content of less than 0.8%, the strength and toughness are lower and therefore the lower limit of the manganese content is set at 0.8%. On the other hand, an excessive manganese content gives an increased hardenability of a HAZ and a considerable amount of bainite or island martensite is formed to deteriorate the toughness of the base steel and the HAZ. Therefore in the present invention, the upper limit of the manganese content is set at 1.8%.

Aluminium is inevitably contained for deoxidation in a killed steel such as that used in this invention, and a total aluminium content of less than 0.01% does not give sufficient deoxidation, leading to deterioration in the toughness of the base steel. Therefore, the lower limit of the aluminium content is set at 0.01% in the present invention. On the other hand, when the total aluminium content exceeds 0.08%, both the HAZ toughness and the toughness of the weld metal are lowered remarkably. Therefore, the upper limit of the total aluminium content is set at 0.08%.

The sulphur content present as an impurity, is limited to not more than 0.015%. High Charpy impact values are required both for the base steel and HAZ in the case of steel pipes for gas pipe lines, but striation phenomena takes place on the impact fracture surface of a CR steel sheet and improves the brittle fracture characteristic but lowers the impact value. In order to improve the impact value, it is particularly effective to maintain the sulphur content to not more than 0.015%. In this case, the lower sulphur content, the more improved the Charpy test toughness, and not more than 0.008% sulphur is desirable for stably obtaining a high level of absorbed energy.

Phosphorus is always contained as an unavoidable impurity in steel, and normally in an amount of not more than 0.03%. Phosphorus is not intentionally added to steel used in this invention, and lower phosphorus contents improve the toughness.

According to a modification of steel composition used in this invention, the basic steel composition may further contain at least one of 0.02 to 0.20% V, not more than 0.6% Cr, not more than 0.6% Cu and not more than 2.5% Ni.

Vanadium can be added for the purpose of improving the strength and toughness of the base steel and for increasing the range of steel sheet thicknesses which can be produced. The addition of vanadium is particularly effective to improve the strength and toughness, especially the strength of a welded portion of the sheet. Thus, in gas pipe lines which are required to have a high level of tensile strength, an increased thickness and simultaneously a satisfactory low-temperature toughness, it is not possible to obtain 40 kg/mm^2 or more of yield strength (equivalent to grades X-65 - X-70) by the addition of molybdenum alone. The rolled austenite grains can be further refined if the molybdenum is added with the presence of vanadium - which has less recrystallization-preventing characteristics than molybdenum - resulting in the rolled austenite grains being elongated more smoothly in the non-crystallization zone, so that the rolled structure can be refined finer. However, with vanadium

contents exceeding 0.20%, precipitated V(CN) is not easily and stably dissolved into solid solution at a heating temperature of 1150°C or lower and the toughness of the base metal as well as of the HAZ deteriorates. Therefore, the upper limit of the vanadium content is set at 0.20%. For maintaining the required strength and toughness, 0.02] or more vanadium is preferred.

Chromium, copper and nickel are added mainly for the purpose of improving the strength and toughness of the base metal, and for increasing the range of possible steel sheet thicknesses for production. The contents are naturally limited to be below certain amounts, but in the low-carbon steel used in this invention without the addition of niobium, the respective upper limits can be higher than those of an ordinary carbon steel.

Chromium, when present in an excessive amount, increases the hardenability of the HAZ and lowers the toughness and the resistance to the welding cracks; therefore, the upper limit of the chromium content is set at 0.6%.

Nickel, up to a certain amount, can improve the strength and toughness of the base metal without having adverse effects on the hardenability and toughness of the HAZ, but a nickel content exceeding 2.5% has an adverse effect on the hardenability and toughness of the HAZ. Therefore, the upper limit of the nickel content is set at 2.5%. Further, in order to improve the stress corrosion resistance in an atmosphere containing hydrogen sulphide, less than 1.0% nickel is desirable.

Copper has a similar effect as nickel and is favourable for corrosion resistance, but copper contents exceeding 0.6% can cause copper-cracks during the sheet rolling, resulting in difficulties in production. Therefore, the upper limit of the copper content is set at 0.6%.

Regarding the lower limits of chromium, nickel and copper, at least 0.1% should be added in order fully to obtain the effect of their addition.

According to further modifications of the present invention, the base steel composition or the modified steel composition defined hereinbefore may further be modified so as to contain one or more of 0.001 to 0.03% REM (rare earth metal), 0.0005 to 0.03% Ca, and 0.004 to 0.03% Ti; when titanium is added, the nitrogen content preferably is limited to lie in the range of 0.001 to 0.009%, and when REM is added, the REM/S ratio preferably is limited to lie in the range of 1.0 to 6.0. By the above further modifications, still greater improvements in toughness can be achieved.

Both REM and Ca are effective at spheroidizing MnS and preventing the elongation of MnS during the CR. This contributes to improving the toughness in the direction perpendicular to the rolling direction, and also prevents ultrasonic testing defects, caused by large, elongated MnS and hydrogen in the steel.

Regarding a content of REM, less than 0.001% produces no practical effect, but more than 0.03% causes both enlargements in the REM-sulphide and also a large amount of REM-oxysulphide, which exists as large size inclusions. A high REM content thus damages not only the toughness but also the cleanliness of the steel sheet. Therefore, in the present invention, the REM content is limited to the range of 0.001 to 0.03%.

Though an REM content is effective at improving and stabilizing the toughness of the steel sheet in co-operation with the sulphur content, the optimum REM content for this purpose is defined by a REM/S ratio ranging from 1.0 to 6.0.

Calcium has a similar effect as REM and its content, when present, is limited to the range from 0.0005 to 0.03%.

Titanium can be added for the purpose of dispersing fine TiN in the steel slab before heating, so as to achieve refinement of the heated austenite grains. In a steel composition containing no niobium, as is used in this invention, the recrystallization takes place down to low temperatures and the recrystallized rolled austenite grains are considerably refined by the molybdenum. If the heated austenite grains are maintained fine, the recystallized rolled austenite grains are refined yet further and the low-temperature toughness is improved even more. To obtain this result, fine TiN particles dispersed in the steel slab will refine the grains on heating and preferably 0.004% or more of TiN particles, not larger than 0.02μ, should be present. However, in an ordinary ingot making process, the solidification and cooling speed is so slow that TiN is apt to precipitate as coarse particles and it is difficult to obtain stably the fine TiN required for the refinement of the heated austenite grains. Therefore, for commercial production, continuous casting is preferred when adding titanium. In this case, however, an excessive titanium content causes precipitation of coarse TiN, and therefore the upper limit of any Ti content is set at 0.03%. On the other hand, a titanium content of less than 0.004% gives no practical effect at refining the heated austenite grains, and therefore the lower limit of the titanium content is set at 0.004%.

Further, in order more effectively to obtain the fine TiN, it is advantageous to limit the nitrogen content in relation to the titanium content so as to lie in the range of from 0.001 to 0.009%. This is because if more titanium is present than the chemical equivalent to nitrogen, TiC is formed and this is harmful to the toughness. Therefore, a titanium content greater than

chemical equivalent to N should be avoided.

Regarding the hot rolling of this invention, a heavy plate rolling mill is most preferably used, but a hot strip mill may instead be used.

By way of illustration of this invention, certain specific Examples of steels and methods of this invention will now be set out together with comparative Examples. Tables 1 to 3 each set out the compositions of both comparative steels and steels of this invention, together with the production processes used and the properties and characteristics of the steels. It can be seen that steel sheets of this invention display excellent properties and especially strength and toughness. The low temperature toughness and the resistance to hydrogen cracking after welding are both especially good.

It will of course be appreciated that the steel sheets produced in accordance with this invention can be used for purposes other than for the manufacture of pipes. The steels are however particularly suited to the manufacture of pipes for pipe-lines to be used in severe climatic conditions owing to the low-temperature toughness.

Table 1

o : Present Invention

Classification	Steels	C	Si	Mn	S	Mo	V	Al	N	Others
		Chemical Composition (%)								
	A-1	0.08	0.26	1.34	0.004	0.26	0.078	0.030	0.0050	Ni 0.25
	A-2	"	"	"	"	"	"	"	"	"
o	A-3	"	"	"	"	"	"	"	"	"
	B-1	0.05	0.10	1.65	0.003	0.28	0.060	0.025	0.0055	Ti 0.014 REM 0.009
	B-2	"	"	"	"	"	"	"	"	"
o	B-3	"	"	"	"	"	"	"	"	"
	C-1	0.09	0.20	1.50	0.003	0.20	–	0.020	0.0060	Ti 0.013 Ca 0.008
	C-2	"	"	"	"	"	–	"	"	"
	C-3	"	"	"	"	"	–	"	"	"
o	C-4	"	"	"	"	"	–	"	"	"
	D-1	0.06	0.20	1.35	0.002	0.30	0.050	0.030	0.0040	Ni 0.80 Cr 0.20 REM 0.011
	D-2	"	"	"	"	"	"	"	"	"
	D-3	"	"	"	"	"	"	"	"	"
o	D-4	"	"	"	"	"	"	"	"	"
	E-1	0.10	0.15	1.45	0.003	0.08	–	0.025	0.0025	–
	E-2	"	"	"	"	"	–	"	"	–
o	E-3	"	"	"	"	"	–	"	"	–
o	E-4	"	"	"	"	"	–	"	"	–
	F-1	0.03	0.15	1.50	0.003	0.25	0.050	0.028	0.0060	Ni 0.20 Cu 0.25 Ti 0.010 REM 0.010
	F-2	"	"	"	"	"	"	"	"	"
o	F-3*1)	"	"	"	"	"	"	"	"	"
o	F-4	"	"	"	"	"	"	"	"	"
	G-1	0.09	0.30	1.43	0.004	0.20	0.040	0.035	0.0055	Nb 0.04
	G-2	"	"	"	"	"	"	"	"	"
	G-3	"	"	"	"	"	"	"	"	"
	G-4	"	"	"	"	"	"	"	"	"
	H-1	0.05	0.25	1.40	0.003	–	0.025	0.025	0.0060	Ni 0.70 Cu 0.26 REM 0.010
	H-2	"	"	"	"	–	"	"	"	"
	H-3	"	"	"	"	–	"	"	"	"
o	I	0.09	0.20	1.50	0.004	0.27	0.080	0.028	0.0070	–
o	J	0.06	0.25	1.55	0.003	0.10	–	0.030	0.0045	Ti 0.012 Ca 0.0008 Cu 0.28 Ni 0.90
o	K	0.08	0.20	1.50	0.003	0.20	–	0.020	0.0055	Ti 0.014

*1) F-3 was subjected to heating at 530°C for 10 minutes to remove the hydrogen immediately after the rolling

Table 1 (continued)

Classi-fication	Steels	Method*2) of Slab Produc-tion	Sheet Production Conditions							Sheet Thick-ness (mm)
			Heat-ing Temp. (°C)	Heat γ Grain Size (ASTM No.)	In the Temp. Range of 900 - 1050°C		Reduction Percen-tage at 900°C or lower (%)	Finish-ing Temp. (°C)	Rolled γ Grain Size (ASTM No.)	
					Number of Passes	Reduction Percentage of Each Pass (in Time Sequence)(%)				
	A-1	IG	1150	3.0	6	3.5, 2.5, 1.5, 2.8, 4.1	60	740	4.5	20
	A-2	"	"	"	6	3.0, 1.0, 6.0, 20.5, 4.5, 4.0	60	730	5.0	20
o	A-3	"	"	"	6	2.5, 4.0, 3.5, 4.5, 3.5, 4.0	59.5	740	6.5	20
	B-1	CC	1150	4.0	5	1.7, 2.8, 3.5, 4.0, 3.5	85	760	5.0	32
	B-2	"	"	"	5	2.5, 1.6, 9.5, 10.0, 4.0	80	760	5.5	32
o	B-3	"	"	"	5	3.0, 4.5, 5.0, 5.5, 3.5	84	750	7.0	32
	C-1	IG	1080	4.5	5	2.5, 1.9, 4.5, 6.5, 5.0	65	740	5.0	16
	C-2	"	"	"	6	5.0, 4.0, 4.5, 4.5, 4.0, 3.5	40	750	5.5	16
	C-3	"	"	"	6	4.5, 5.0, 4.5, 5.0, 4.5, 4.0	70	840	5.9	16
o	C-4	"	"	"	6	5.0, 4.0, 4.5, 5.0, 4.5, 4.5	70	750	7.0	16
	D-1	IG	1150	3.0	7	2.5, 1.0, 5.0, 4.5, 3.0, 2.5, 4.5	65	780	4.0	20
	D-2	"	"	"	7	8.5, 9.0, 1.5, 4.5, 2.5, 4.0, 1.8	60	760	5.0	20
	D-3	"	1250	2.0	8	3.5, 4.5, 5.0, 4.5, 4.0, 4.5, 4.0, 4.5	75	750	4.5	20
o	D-4	"	1150	3.0	9	5.5, 2.5, 3.5, 5.0, 4.0, 5.5, 4.0	70	760	7.0	20
	E-1	CC	1150	3.5	6	2.5, 1.5, 6.0, 3.5, 4.0, 5.0	40	835	4.5	26
	E-2	"	1200	2.5	6	5.0, 4.5, 4.0, 6.0, 6.5, 3.5	60	760	5.5	26
o	E-3	CC	1150	3.5	6	2.5, 2.5, 4.0, 5.5, 4.5, 5.0	70	750	7.0	26
o	E-4	"	1150	3.5	6	4.5, 5.0, 4.5, 4.5, 4.0, 60	80	750	7.0	26

Table 1 (continued)

Classification	Steels	Method*2) of Slab Production	Sheet Production Conditions: Heating Temp. (°C)	Heat γ Grain Size (ASTM No.)	In the Temp. Range of 900 - 1050°C: Number of Passes	Reduction Percentage of Each Pass (in Time Sequence) (%)	Reduction Percentage at 900°C or lower (%)	Finishing Temp. (°C)	Rolled γ Grain Size (ASTM No.)	Sheet Thickness (mm)
	F-1	IG	1150	4.5	8	10.0, 5.0, 1.9, 3.5, 2.0, 4.5, 4.0, 3.5	70	760	5.5	20
	F-2	”	”	”	7	3.5, 4.0, 6.5, 8.0, 3.5, 10.0, 3.5, 4.0	40	780	5.9	20
o	F-3*1)	”	”	”	9	3.6, 4.0, 5.0, 4.5, 4.0, 3.0, 4.0, 3.5, 3.8	60	760	7.0	20
o	F-4	”	”	”	8	4.5, 4.0, 3.0, 4.5, 4.5, 4.5, 4.0, 5.0	70	780	8.0	20
	G-1	IG	1150	2.5	6	3.5, 1.8, 4.5, 5.0, 3.0, 3.5	70	720	4.5	32
	G-2	”	”	”	6	1.9, 6.0, 8.0, 3.5, 4.0, 3.5	75	750	4.5	32
	G-3	”	”	”	6	4.5, 5.0, 4.0, 3.5, 4.0, 4.5	60	740	5.5	32
	G-4	”	”	”	5	1.0, 1.5, 8.5, 10.5, 0.5	75	745	5.8	32
	H-1	CC	1080	2.8	7	2.8, 1.6, 4.0, 3.5, 2.5, 3.5, 3.0	70	780	4.0	16
	H-2	”	”	”	7	4.0, 3.0, 4.5, 4.0, 5.0, 5.0, 4.5	70	765	5.5	16
	H-3	”	”	”	6	1.5, 8.0, 1.5, 3.5, 8.0, 2.5	75	760	5.6	16
o	I	IG	1150	3.0	5	2.5, 3.0, 5.0, 6.0, 4.5	70	730	6.0	20
o	J	CC	1150	4.0	5	2.0, 8.0, 5.0, 4.0, 3.5	65	740	6.5	13.7
o	K	CC	1000	4.5	6	5.0, 4.0, 3.5, 4.5, 6.0, 4.0	70	740	6.5	16

*2) CC: Continuous Casting Process; IG: Ingot Making Process

Table 1 (continued)

Classification	Steels	Properties of Base Metal *3)						Impact Absorbed Energy in Welded Portion −40°C, 2mmV Notch Charpy (kg-m)	Number of Cross Sectional Cracks in Hydrogen Cracking Resistance Test (mm)
		Tensile Properties			2mmV Charpy Impact Properties		DWTT *4) 85% SATT (°C)		
		Yield Point (kg/mm²)	Tensile Strength (kg/mm²)	Elongation (%)	vE-60°C (kg-m)	vTrs (°C)			
	A-1	46.5	58.1	42	4.8	− 40	− 5	8.0	3
	A-2	47.1	59.0	42	6.1	− 50	−10	7.0	2
o	A-3	49.1	61.5	41	14.2	−105	−40	8.5	2
	B-1	48.1	61.0	48	6.0	− 50	+10	14.1	2
	B-2	49.0	61.5	46	7.1	− 60	0	13.8	2
o	B-3	51.5	63.0	49	18.1	−125	−40	14.9	1
	C-1	46.1	57.5	36	6.0	− 30	−10	9.5	2
	C-2	47.1	58.1	38	6.5	− 40	−21	9.8	3
	C-3	41.5	54.6	36	6.8	− 50	−20	9.0	2
o	C-4	51.2	60.1	37	14.5	−100	−65	10.6	1
	D-1	50.6	61.6	40	4.0	− 40	− 5	12.1	3
	D-2	50.4	61.9	41	3.8	− 50	−10	12.0	3
	D-3	51.2	62.1	42	6.0	− 70	−25	11.4	4
o	D-4	51.8	63.0	42	10.5	−120	−50	11.8	2
	E-2	46.1	58.1	36	5.1	− 50	0	8.5	6
o	E-3	48.9	59.0	38	11.6	−105	−40	9.0	3
o	E-4	48.0	58.9	39	12.8	−100	−45	9.5	2
	F-1	50.1	60.1	38	9.1	− 70	−20	15.1	2
	F-2	50.6	60.0	39	9.8	− 75	−20	16.2	1
o	F-3*1)	53.5	62.5	40	24.5	−140	−70	16.0	0
o	F-4	51.9	64.1	40	22.5	−120	−60	16.4	1

Table 1 (continued)

Classification	Steels	Properties of Base Metal *3)						Impact Absorbed Energy in Welded Portion −40°C, 2mmV Notch Charpy (kg-m)	Number of Cross Sectional Cracks in Hydrogen Cracking Resistance Test (mm)
		Tensile Properties			2mmV Charpy Impact Properties		DWTT *4) 85% SATT (°C)		
		Yield Point (kg/mm²)	Tensile Strength (kg/mm²)	Elongation (%)	vE-60°C (kg-m)	vTrs (°C)			
	G-1	50.5	62.5	43	4.8	− 45	+ 5	4.5	5
	G-2	51.0	62.8	46	4.9	− 50	+ 2	4.8	8
	G-3	51.1	63.4	44	5.1	− 50	0	4.4	6
	G-4	52.5	64.1	44	5.9	− 55	−10	4.6	3
	H-1	48.5	59.1	38	4.1	− 55	− 8	6.2	4
	H-2	49.1	60.2	36	4.4	− 60	−10	6.4	2
	H-3	50.0	61.4	38	5.1	− 65	−15	6.9	5
o	I	46.0	57.5	42	10.1	− 95	−35	8.2	2
o	J	46.5	56.4	38	20.6	−105	−45	15.0	0
o	K	43.5	54.6	43	15.1	−100	−40	14.0	1

*3) Properties of the base metal are expressed by values in the direction perpendicular to the final rolling direction.

*4) 85% ductility-fracture transition temperature (API standard), namely the temperature at which the ductility-fracture ratio is 85%. Refer to Fig. 7 and Fig. 8.

Table 2

Classification	Steels	Chemical Composition (%)									Method *1) of Slab Production
		C	Si	Mn	S	Mo	Nb	V	Al	N	
Present Invention	1	0.09	0.37	1.32	0.004	0.12	–	–	0.022	0.0088	CC
	2	0.04	0.25	1.10	0.003	0.10	–	–	0.028	0.0042	IG
	3	0.08	0.16	1.22	0.009	0.17	–	0.04	0.016	0.0097	IG
	4	0.10	0.33	1.45	0.005	0.28	–	0.05	0.033	0.0079	IG
	5	0.03	0.22	1.65	0.005	0.21	–	0.07	0.019	0.0068	CC
Comparison	6	0.09	0.37	1.32	0.004	0.12	–	–	0.022	0.0088	CC
	7	"	"	"	"	"	–	–	"	"	CC
	8	"	"	"	"	"	–	–	"	"	CC
	9	0.15	0.21	1.40	0.008	0.45	0.02	–	0.019	0.0090	IG
	10	0.10	0.33	1.45	0.005	0.28	–	0.05	0.033	0.0079	IG
	11	"	"	"	"	"	–	"	"	"	IG
	12	0.06	0.32	1.80	0.007	0.35	0.04	0.15	0.022	0.0120	IG

*1) CC: Continuous Casting Process; IG: Ingot-making Process

Table 2 (continued)

Classification	Steels	Sheet Production Conditions: Heating Temp. (°C)	Heated γ Grain Size (ASTM No.)	In the Temp. Range of 900–1050°C: Number of Passes	In the Temp. Range of 900–1050°C: Reduction Percentage of Each Pass (in Time Sequence) (%)	Reduction Percentage at 900°C or lower (%)	Finishing Temp. (°C)	Rolled γ Grain Size (ASTM No.)	Sheet Thickness (mm)
Present Invention	1	1150	3.0	4	8.0, 8.5, 9.0, 15.0	75	735	7.0	16
	2	1150	3.0	7	9.0, 8.0, 10.0, 15.0, 8.5, 9.5, 10.0	75	800	6.5	8
	3	1150	3.5	5	9.5, 9.0, 12.0, 13.0, 12.5	70	730	7.5	16
	4	1100	4.0	4	12.0, 10.0, 15.0, 13.5	65	745	8.0	20
	5	1150	3.0	6	8.0, 9.5, 8.0, 15.0, 13.0, 10.5	75	750	7.5	20
Comparison	6	1300	2.0	4	10.0, 12.5, 10.0, 15.0	60	740	5.0	16
	7	1150	3.0	0	—	70	735	5.5	16
	8	1150	3.0	5	3.0, 4.0, 4.5, 5.0, 4.5	40	840	4.5	16
	9	1250	0	3	8.0, 13.5, 15.0	60	750	3.5	20
	10	1150	3.0	4	10.0, 12.5, 12.0, 15.0	30	850	4.5	16
	11	"	3.0	1	20.0	60	760	5.0	16
	12	1250	−1.0	6	10.0, 11.5, 10.5, 15.0, 20.0, 15.0	70	750	4.0	20

Table 2 (continued)

Classification	Steels	Properties of Base Metal *2)					
		Tensile Properties			2mmV Charpy Impact Properties		DWTT *3) 85% SATT (°C)
		Yield Point (kg/mm²)	Tensile Strength (kg/mm²)	Elongation (%)	vE-60°C (kg-m)	vTrs (°C)	
Present Invention	1	46.2	52.5	42	13.8	−102	−45
	2	43.9	50.1	38	14.2	−105	−60
	3	52.1	58.5	37	10.8	−121	−65
	4	51.1	61.8	40	10.5	−110	−60
	5	48.1	61.8	42	20.3	−107	−60
Comparison	6	45.5	51.9	40	5.3	− 60	−15
	7	46.9	52.3	39	5.8	− 75	−20
	8	36.3	46.9	54	2.9	− 52	0
	9	51.3	68.2	37	1.9	− 46	+10
	10	42.5	58.8	45	5.7	− 41	+10
	11	52.2	61.8	37	7.4	− 82	−20
	12	56.3	67.4	38	8.2	− 83	−15

*2) Properties of the base metal are expressed by values in the direction perpendicular to the final direction.

*3) 85% ductility-fracture transition temperature (API standard), namely the temperature at which the ductility-fracture ratio is 85%. Refer to Fig. 7 and Fig. 8.

Table 3

Classification	Steels	Chemical Composition (%) C	Si	Mn	S	Mo	Nb	V	Cr	Cu	Ni	REM	Ca	Ti	N	REM/S	Method of Slab Production *1)
Present Invention	1	0.12	0.33	1.28	0.010	0.21	–	–	0.22	–	0.45	–	–	–	0.0048	–	IG
	2	0.07	0.29	1.52	0.002	0.18	–	0.05	–	0.25	0.21	–	–	–	0.0079	–	IG
	3	0.06	0.25	1.50	0.003	0.21	–	0.07	–	–	0.28	0.012	–	–	0.0055	4	CC
	4	0.10	0.28	1.44	0.004	0.22	–	0.07	–	0.25	–	0.008	0.001	–	0.0080	2	IG
	5	0.09	0.15	1.23	0.005	0.13	–	–	0.26	–	–	–	–	0.018	0.0096	–	CC
	6*2)	0.04	0.25	1.49	0.002	0.27	–	0.06	–	–	1.34	–	–	0.019	0.0077	–	CC
	7	0.07	0.13	1.21	0.005	0.16	–	0.04	–	–	–	–	0.003	0.016	0.0098	–	CC
	8	0.07	0.13	1.21	0.005	0.16	–	0.04	–	–	–	–	0.003	0.016	0.0098	–	CC
	9	0.06	0.20	1.65	0.003	0.20	–	–	–	–	–	–	–	0.012	0.0035	–	CC
Comparison	10	0.13	0.22	1.58	0.012	0.30	–	0.09	–	0.25	0.22	–	–	0.042	0.0081	–	CC
	11	0.06	0.32	1.55	0.003	0.45	–	0.07	–	–	0.98	0.009	–	–	0.0065	3	IG
	12	0.09	0.31	1.22	0.004	0.04	0.05	0.06	–	–	0.25	–	–	0.013	0.0071	–	IG

*1) CC: Continuous Casting Process; IG: Ingot-Making Process

*2) Subjected to heating at 530°C for 10 minutes to remove the hydrogen immediately after the rolling.

Table 3 (continued)

Classification	Steels	Sheet Production Conditions: Heating Temp. (°C)	Heated γ Grain Size (ASTM No.)	In the Temp. Range of 900 – 1050°C: Number of Passes	In the Temp. Range of 900 – 1050°C: Reduction Percentage of Each Pass (in Time Sequence) (%)	Reduction Percentage at 900°C or lower (%)	Finishing Temp. (°C)	Rolled γ Grain Size (ASTM No.)	Sheet Thickness (mm)
Present Invention	1	1050	6.0	4	15.0, 15.0, 15.0, 15.0	70	740	8.5	16
	2	1150	3.0	3	15.0, 10.0, 20.0	75	730	7.5	12
	3	1150	3.0	6	10.0, 12.0, 12.5, 15.0, 15.0, 15.0	70	720	7.5	20
	4	1150	2.5	6	9., 8.0, 15.0, 10.0, 12.0, 10.5	75	720	7.0	18
	5	1150	5.5	5	10.0, 15.0, 15.5, 20.0, 15.0	75	740	8.5	16
	6*2)	1100	7.0	7	10.0, 15.0, 14.0, 15.0, 13.0 12.0, 10.0	75	750	9.0	30
	7	1150	6.0	7	10.0, 12.5, 15.0, 14.0, 10.0 15.0, 20.0	80	690	8.5	16
	8	1150	6.0	5	3.0, 6.0, 8.0, 8.0, 4.0	70	720	8.0	16
	9	980	7.0	5	4.0, 8.0, 10.0, 3.0, 8.0	65	720	8.5	26
Comparison	10	1150	3.5	4	10.0, 15.0, 13.0, 12.0	75	750	7.0	15
	11	1250	−1.0	5	15.0, 12.0, 15.0, 10.0, 10.5	70	740	7.0	30
	12	1150	3.5	5	9.5, 8.0, 10.0, 20.0, 15.0	75	740	5.0	15

*2) Subjected to heating at 530°C for 10 minutes to remove the hydrogen immediately after the rolling.

Table 3 (continued)

Classification	Steels	Properties of Base Metal *3)					
		Tensile Properties			2mmV Charpy Impact Properties		DWTT 85% SATT (°C)
		Yield Point (kg/mm^2)	Tensile Strength (kg/mm^2)	Elongation (%)	vE-60°C (kg-m)	vTrs (°C)	
Present Invention	1	48.0	56.3	40	12.3	−112	−60
	2	54.9	65.7	42	20.8	−128	−65
	3	50.3	62.9	44	20.3	−115	−60
	4	53.4	64.1	43	18.6	−106	−55
	5	47.1	54.4	41	16.2	−129	−70
	6	52.8	59.9	51	24.8	−120	−65
	7	52.8	64.4	42	15.9	−123	−75
	8	50.6	62.5	43	13.8	−118	−65
	9	48.7	57.1	44	21.6	−135	−75
Comparison	10	57.9	70.1	35	2.1	− 52	+10
	11	51.3	64.2	47	11.2	− 80	−25
	12	52.1	58.1	40	8.0	− 81	−20

*3) Properties of the base metal are expressed by values in the direction perpendicular to the final rolling direction.

WHAT WE CLAIM IS:

1. In a method of producing a steel sheet (as defined herein) the steps of heating a steel slab containing (by weight) o.01 to 0.13%C, 0.05 to 0.8% Si, 0.8 to 1.8% Mn, 0.01 to 0.08% total Al, 0.08 to 0.40% Mo, and not more than 0.015% S with the balance being iron and unavoidable impurities to a temperature not higher than 1150°C, and hot rolling the heated steel slab, the hot rolling including at least three passes with a minimum reduction percentage of not less than 2% in each rolling pass within the temperature range of 900 to 1050°C, and the total hot rolling reduction percentage at 900°C or lower being not less than 50%, and the finishing temperature of the hot rolling being not higher than 820°C.

2. A method according to Claim 1, in which the steel slab further contains 0.02 to 0.2% V.

3. A method according to Claim 1 or Claim 2, in which the steel slab further contains at least one of 0.001 to 0.03% REM, 0.0005 to 0.03% Ca, and 0.004 to 0.03% Ti, with the limitation that where Ti is contained the steel has a N content of from 0.001 to 0.009%, and when rem is contained, the ratio of REM/S is in the range of from 1.0 to 6.0.

4. A method according to any of Claims 1 to 3, in which the steel slab further contains at least one of not more than 0.6% Cr, not more than 0.6% Cu, and not more than 2.5% Ni.

5. A method according to any of Claims 1 to 4 in which the reduction percentage of said three of more passes is more than 5% per pass.

6. A method according to any of the preceding Claims, in which the slab is heated to a temperature in the range of from 1050 to 1150°C.

7. A method according to any of the preceding Claims, in which the slab is held at the heating temperature for not more than 2 hours.

8. A method according to any of the preceding Claims, in which the heated slab is initially hot rolled at a temperature above 1050°C whereafter said three or more passes in the temperature range of 900° to 1050°C are performed.

9. A method according to any of the preceding Claims, in which the C content of the steel slab is not more than 0.1%.

10. A method according to Claim 1 and substantially as hereinbefore described, with reference to the accompanying drawings.

11. A method according to Claim 1 and substantially as described in the Examples of Table 1 or Table 2 or Table 3, set out hereinbefore.

12. Steel sheet (as defined herein) whenever produced by a method according to any of Claims 1 to 11.

For the Applicants:
SANDERSON AND CO.,
Chartered Patent Agents,
97 High Street,
Colchester,
Essex.

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FIG.1

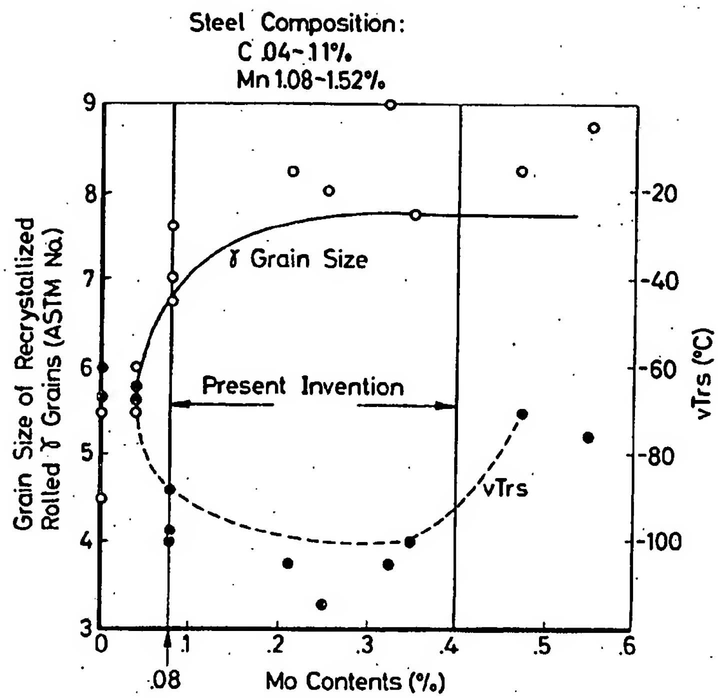

Steel Composition:
C .04~.11%
Mn 1.08~1.52%
Grain Size of Recrystallized Rolled γ Grains (ASTM No.)
vTrs (°C)
γ Grain Size
Present Invention
vTrs
Mo Contents (%)
.08
9
8
7
6
5
4
3
0
.1
.2
.3
.4
.5
.6
-20
-40
-60
-80
-100

FIG. 2

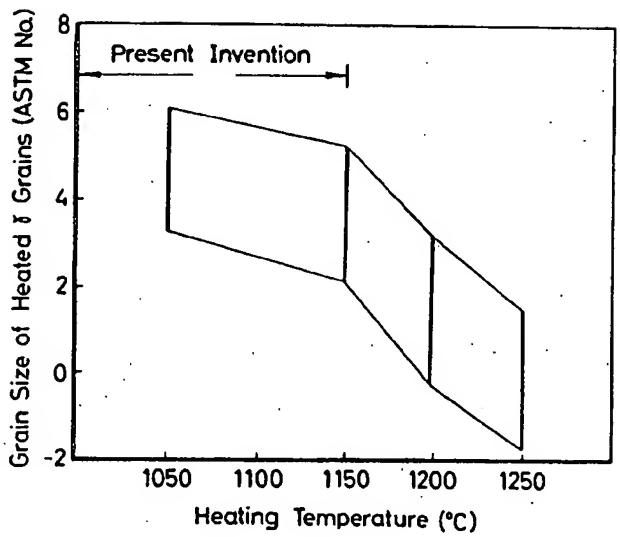

Present Invention
Grain Size of Heated γ Grains (ASTM No.)
8
6
4
2
0
-2
1050
1100
1150
1200
1250
Heating Temperature (°C)

FIG.3

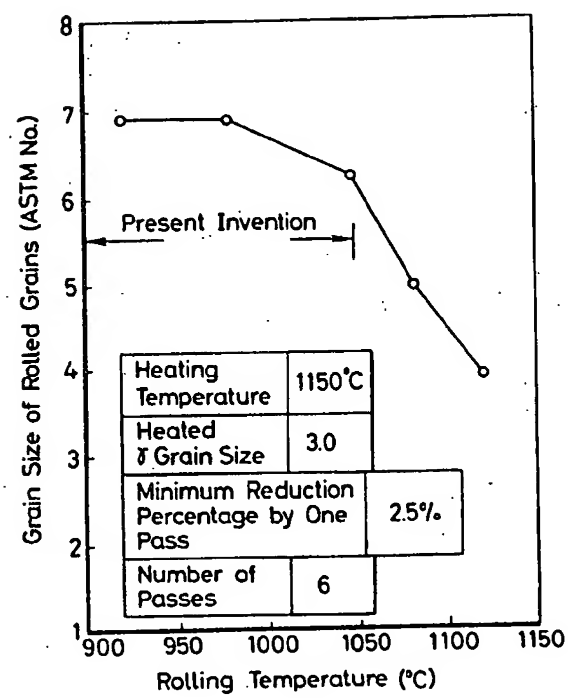

8
7
6
5
4
3
2
1
Grain Size of Rolled Grains (ASTM No.)
Present Invention
Heating Temperature
1150°C
Heated γ Grain Size
3.0
Minimum Reduction Percentage by One Pass
2.5%
Number of Passes
6
900
950
1000
1050
1100
1150
Rolling Temperature (°C)

FIG.4

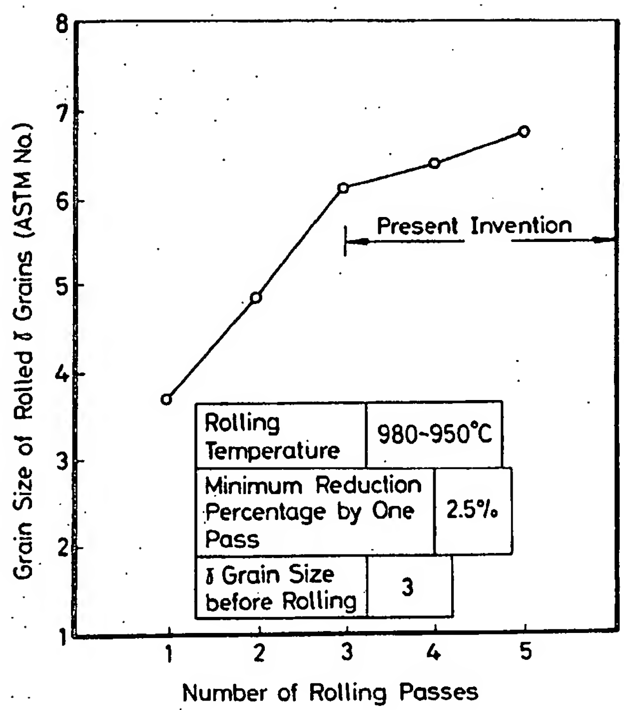
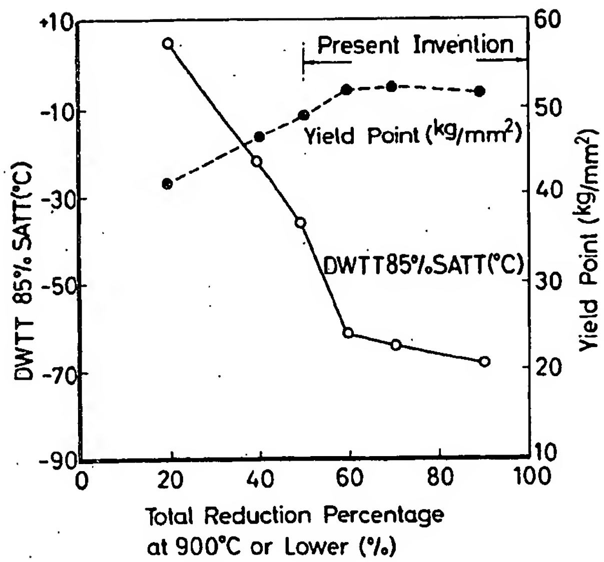

Grain Size of Rolled δ Grains (ASTM No.)
8
7
6
5
4
3
2
1
Present Invention
Rolling Temperature
980~950°C
Minimum Reduction Percentage by One Pass
2.5%
δ Grain Size before Rolling
3
1
2
3
4
5
Number of Rolling Passes

FIG.5


+10
-10
-30
-50
-70
-90
60
50
40
30
20
10
0
20
40
60
80
100
Present Invention
Yield Point (kg/mm²)
DWTT85%SATT(°C)
DWTT 85% SATT(°C)
Yield Point (kg/mm²)
Total Reduction Percentage
at 900°C or Lower (%)

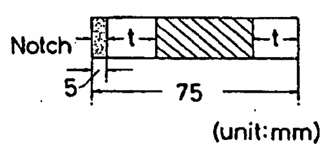

Notch
t
t
5
75
(unit:mm)

FIG.8

FIG.6

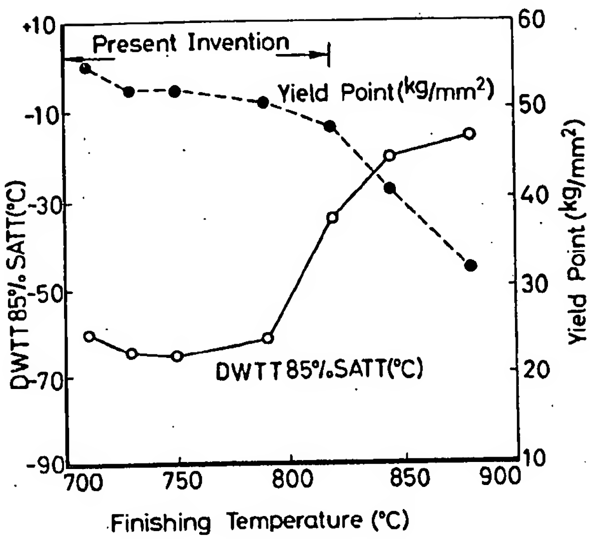
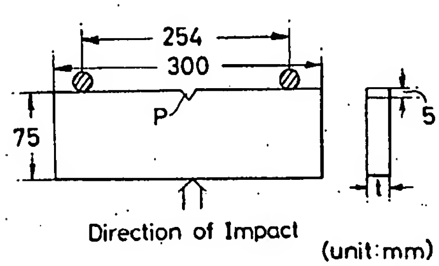

Present Invention
Yield Point (kg/mm²)
DWTT 85%.SATT(°C)
DWTT 85%.SATT(°C)
Yield Point (kg/mm²)
Finishing Temperature (°C)
+10
-10
-30
-50
-70
-90
700
750
800
850
900
60
50
40
30
20
10

FIG.7


254
300
P
75
5
t
Direction of Impact
(unit: mm)